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(54) MANUFACTURE OF THIN STRIP STEEL SHEET

(57)Abstract:

PURPOSE: To eliminate a hot rolling stage and to obtain a steel sheet excellent in strength, toughness or the like from a continuously cast slab by subjecting a cast steel having specified content of C, Si, Mn, P, etc., to primary working under specified conditions and thereafter executing primary cooling, reheating, secondary cooling or the like.

CONSTITUTION: A molten steel contg., by weight, 0.01 to 0.2% C, $\leq 3\%$ Si, 0.5 to 3% Mn, $\leq 0.1\%$ P, 0.003 to 0.05% S and 0.01 to 0.1% Al is continuously cast. This cast steel is non-worked in an austenitic area of the Ae3 transformation point or above, or is subjected to primary working at $\leq 80\%$ total rolling reduction. Next, the steel is cooled from the Ae3 transformation point at $\geq 5^\circ\text{C/s}$ average cooling rate, and its temp. is regulated to less than the higher one between the temp. 1 shown by a formula or 600°C to form a transgranular ferritic structure. Then the steel is again heated to the Ae3 transformation point or above into an austenitic single phase structure and is again subjected to secondary cooling. Then, a thin strip steel sheet having small anisotropy and constituted of a fine polygonal ferritic structure can be obtd.

$$T_1 = A \cdot e_1 - 2000 \times C \text{ (}^\circ\text{C)}$$

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SPECIFICATIONS

1. Title of the Invention

Method of Continuously Casting Thin Steel Sheet

2. Claims

1. A method of continuously casting thin steel strip in which molten steel having the
15 composition

C: 0.01 wt% to 0.2 wt%

Si: Not more than 3.0 wt%

Mn: 0.5 wt% to 3.0 wt%

P: Not more than 0.10 wt%

20 S: 0.003 wt% to 0.05 wt%

Al: 0.01 wt% to 0.1 wt%

with the balance consisting of iron and the inevitable impurities is cast by continuous casting whereupon the steel being in the austenite region at or above the A_{e_3} transformation point the steel is either not worked or alternatively is subjected to primary working to a reduction of not more than 80%, whereupon such steel is subjected to primary cooling at a mean rate of cooling of not less than 5°C/second from the A_{e_3} transformation point to either 600°C or alternatively that temperature T_1 which satisfies the requirements of formula (1), whichever temperature is the higher, whereby an intragranular ferrite structure is formed, and when such transformation has been fully completed, such steel is again heated to not less than the A_{e_3} transformation point whereby a single phase austenite structure is formed, whereupon such steel is again subjected to secondary cooling, and thus provides a thin steel sheet with a polygonal ferrite microstructure and with very low anisotropy.

$$T_1 = A_{e_3} - 2000 \times C \text{ (wt\%)} \quad (1)$$

2. A method of manufacturing thin steel sheet as described in paragraph 1 of these Claims, wherein such steel possesses a normal composition containing one or a plurality selected from not more than 1.0% of Nb, Ti, V, Ni, Cr, Mo, and B.

3. Detailed Description of the Invention

Area of application within industry

The present invention relates to a method for the manufacture of steel sheet, and more particularly to a method of manufacture of steel sheet in which the continuously cast ingot directly after casting, either without undergoing the hot rolling stage at all or alternatively undergoing limited hot rolling, possesses excellent mechanical properties such as the balance

between strength and ductility and so forth which are required of conventional hot rolled steel sheet.

Prior art

The significant developments which have occurred over recent years in continuous casting machines have made it possible to produce thinner cast ingot, whereby the omission or curtailment of the conventional hot rolling process has been advanced. Moreover, such reduction in thickness is of interest for its ability to reduce the costs incurred in the hot rolling process.

However, in such processes which manufacture sheet from cast ingot which is close to the final thickness of such sheet, the absence of the high degree of reduction associated with the hot rolling process means that, generally, the ferrite microstructure becomes large and coarse and such steel sheet suffers the deficiency that the mechanical properties thereof are unsatisfactory. Thus the rate of reduction in the conventional hot rolling process is said to be not less than 80%. (Yoshio Hashimoto: *Tetsu-to Hagane*, 72 (1986), p. 2263)

The reason for the coarsening of the structure which was referred to in the foregoing is that when the transformation from an austenite structure to a ferrite structure occurs, the austenite has a coarse grain structure and the area of the grain boundaries is small, and there are few defects such as deformation zones, such that there are few ferrite nucleation sites.

In recent years, efforts to overcome this problem have been concentrated on the use of oxides and so forth as ferrite transformation nuclei from within the austenite grains, by employing the ferrite within the austenite grains to vigorously drive such transformation, in the case of steel plate in particular. (See for example Japanese Patent No. 213322 of 1986 [61-213322].) However, this method is difficult to employ because of the melt conditions and component control which are required in order to control the microstructure in such a manner that the

oxides and precipitates and so forth are uniformly dispersed, while it is difficult to achieve a uniform microstructure because of the competition between the ferrite side plates that are formed from the austenite grain boundaries and the bainite and so forth during transformation. Moreover, the intragranular ferrite structures readily form highly acicular Widmanstätten ferrite microstructure, and while this is very effective in imparting toughness, it is not very effective in imparting such mechanical properties as ductility and fatigue strength. Moreover, the greatest problem associated with this technique is that the aforementioned structure is encouraged to grow in an uneven manner when the austenite is rolled, and localized occurrences of such phenomena are highly detrimental to such important characteristics as ductility.

However, in-line reheating treatment has been proposed by for example Japanese Patent No.99630 of 1986 (61-99630) and Japanese Patent No. 62822 of 1988 (63-62822) and so forth as a further approach to the problem of the coarsening of the grain structure in thin cast ingot. However, these methods assume cold rolled steels for press forming and with excellent deep drawing characteristics and are under consideration as processes to replace hot rolling in order to obtain stock for cold rolling.

For these reasons, attention has been concentrated on whether the reverse transformation from ferrite to austenite through in-line reheating would be effective in the randomization of the characteristic (100) cluster structures in the as-cast structure, but no reference is made at all to the mechanical properties of the steel sheet at the equivalent of the completion of the hot rolling process.

Problems addressed by the present invention

It is an object of the present invention to establish a method for the manufacture directly from continuously cast ingot, and either without the hot rolling process being employed or

alternatively with the hot rolling process being curtailed, of steel sheet which possesses excellent mechanical properties and which possesses the balance between strength and ductility and so forth which is possessed by conventional hot rolled steel sheet.

Means employed in order to overcome such problems

5 The metallurgical microstructural characteristics that are observed in common in all hot rolled steel sheet, which possesses excellent mechanical properties such as balance between strength and ductility and fatigue strength and so forth, are that such steels possess very fine and uniform microstructures, and that the individual crystal grains are polygonal. Thus, in the simplified process which uses thin cast ingot as the starting point, the desired mechanical
10 properties may be achieved by ensuring that the final microstructure which is derived from the transformation of the austenite microstructure is predominantly of fine-grained polygonal ferrite.

As a result of careful research and in the light of the foregoing, the inventors of the present invention have reached the conclusion that it is not possible to generate a completely po-
15 lygonal ferrite structure only through the transformation of the ferrite structures at the boundaries only as has been studied under the prior art.

Moreover, in the process by which the reverse transformation is imposed, the effect of increasing the fineness of the grain size produced by the repeated heat treatment is strongly affected by the previous microstructure, and more particularly, it has been confirmed that,
20 even when a previous microstructure, which is a coarse ferrite structure such as is obtained by normal solidification, is transformed into austenite by a single reverse transformation, and then undergoes transformation to ferrite through a repeated cooling, there is relatively little change in the diameter of the grains.

The latter phenomenon has been confirmed by Japanese Patent No. 115654 of 1988 (63-115654), and the inventors of that invention require the repeated heat treatment to be performed not less than twice.

However, in order to overcome such various deficiencies, the inventors of the present invention have discovered through various experiments a process which provides a fine-grained polygonal ferrite microstructure through the application of only one reverse transformation.

The basic reason for this is that through the thermal history that is illustrated in Figure 1, when the reverse transformation from ferrite to austenite occurs, the initial ferrite structure becomes an intragranular ferrite structure within the microstructure.

In other words, the inventors of the present invention control the components and the primary cooling conditions at the initial ferrite transformation, whereby the coarse austenite structure which forms after solidification is rendered into a fine microstructure through the intragranular ferrite within the microstructure, whereupon the reverse transformation is imposed, whereby such features are found wherever the fine-grained austenite structure is formed, and as a result, the final secondary cooling as envisaged by the present invention provides a fine-grained ferrite microstructure, and such structure is irregular and polygonal.

This mechanism is not yet understood in detail, but it is believed that the intragranular ferrite structures within the microstructure hold a very large number of austenite nucleation sites when the austenite transformation occurs, and the orientations of the intragranular ferrite within the microstructure are all different from each other, and hence, when the reverse transformation occurs, each of the austenite grains which form from nuclei at the ferrite grain boundaries also possess varying orientations, and thus the fine-grained microstructure is maintained even at the completion of the formation of the austenite.

Moreover, the imposition of the reverse transformation, as envisaged by the present invention, is able to overcome the problem of the structural irregularity associated with the intragranular ferrite within the microstructure, and hence the conditions in the prior art, which are intended to reduce such irregularity and which apply to the composition of the steel and to the reduction, and which restrict the manufacturing process, may be moderated.

Thus the present invention relates to a method of continuously casting thin steel strip in which molten steel having the composition C: 0.01 wt% to 0.2 wt%, Si: Not more than 3.0 wt%, Mn: 0.5 wt% to 3.0 wt%, P: Not more than 0.10 wt%, S: 0.003 wt% to 0.05 wt%, Al: 0.01 wt% to 0.1 wt%, with the balance consisting of iron and the inevitable impurities is cast by continuous casting whereupon the steel being in the austenite region at or above the A_{e_3} transformation point the steel is either not worked or alternatively is subjected to primary working to a reduction of not more than 80%, whereupon such steel is subjected to primary cooling at a mean rate of cooling of not less than 5°C/second from the A_{e_3} transformation point to either 600°C or alternatively that temperature T_1 which satisfies the requirements of formula (1), whichever temperature is the higher, whereby an intragranular ferrite structure is formed, and when such transformation has been fully completed, such steel is again heated to not less than the A_{e_3} transformation point whereby a single phase austenite structure is formed, whereupon such steel is again subjected to secondary cooling, and thus provides a thin steel sheet with a polygonal ferrite microstructure and with very low anisotropy, and which possesses excellent balance between strength and ductility, toughness, and fatigue strength.

$$T_1 = A_{e_3} - 2000 \times C \text{ (wt\%)} \quad (1)$$

Steels which possess a normal composition containing one or a plurality selected from not more than 1.0% of Nb, Ti, V, Ni, Cr, Mo, and B may also be employed.

First, the reasons for the restrictions upon the chemical compositions as envisaged by the present invention are explained.

5 C is the most important element in the determination of the form of the microstructure when the transformation from austenite to ferrite occurs, and the lower limit for C in the method envisaged by the present invention is 0.01%, in order to obtain a stable Widmanstätten ferrite structure as the initial transformation microstructure. Moreover the upper limit for C is made 0.2% in order to prevent any degradation of the welding characteristics of the steel.

10 Si is added in order to improve the strength of the steel without adversely affecting the strength-ductility balance thereof, while Si also acts to raise the transformation point and to promote the precipitation of ferrite. However, the addition of excessive amounts of Si would cause adverse effects to the ductility and surface appearance of the steel, and hence an upper limit of 3% of Si is imposed.

15 Mn is an element which is added in order to impart great strength to steel, similarly to C, and is also an indispensable element in the formation of the intragranular ferrite structure within the microstructure. Thus the Mn forms MnS with the S, which will be discussed subsequently, and is preferentially precipitated upon the oxides which are formed by the oxide-forming elements in the steel such as Al, Si and Mn, and forms effective nuclei for the precipitation
20 of ferrite and promotes the ferrite transformation within the grain microstructure. For these reasons, the lower limit for the amount of Mn to be added is set at 0.5% at which level such effects become apparent, and the upper limit is set at 3.0% in order to hold down the costs of control of the structure in the manufacture of the steel and also in order to prevent any degradation of the workability of the steel.

P is an element which is added in order to strengthen the steel, but if excessive amounts of P are added, the ductility and welding characteristics of the steel are affected adversely, and hence an upper limit of 0.10% of P is set.

5 S, similarly to Mn, plays an important role in the method envisaged by the present invention in that it forms MnS. However, if excessive amounts of S are added, hot cracking of the steel occurs, and hence an upper limit of 0.05% of S is set, while a lower limit of 0.003% of S is set having regard to the problems of the increasing costs of desulphurizing the steel and so forth.

10 Al is necessary for the deoxygenation of the steel, and a lower limit of 0.01% of Al is set for the present invention. However, the addition of excessive amounts of Al would cause the cost of the steel to increase and such excess would remain as inclusions in the steel which would cause cracking and other problems, and hence the upper limit for Al is set at 0.1%.

15 Nb, Ti, V, Ni, Cr, Mo and B are added to steel for the purpose of strengthening the steel. Ti and V in particular are highly effective in promoting the ferrite transformation within the grains of the microstructure. For these reasons, the present invention envisages the addition to the steel of not more than 1.0% of one or a plurality selected from these elements.

The following is a description of the method of manufacture envisaged by the present invention.

20 In the method of manufacture envisaged by the present invention, steel having the composition described in the foregoing is continuously cast whereupon primary working to a total reduction of not more than 80% is applied to the steel in an as-cast state or alternatively at a temperature of not less than the A_{e3} transformation point, whereupon the initial ferrite transformation takes place. The upper limit to the amount of working at this stage is applied because any greater amount of working would be no different from the conventional hot

rolling process, because the metallurgical structure achieved would be no different from that achieved with the conventional hot rolling process, and because equipment and manufacturing costs would be no different from those incurred through the employment of the conventional hot rolling process. However, any light reduction performed for the purpose of improving the surface condition of the cast ingot must fall within such limit to the amount of reduction.

Next, the primary cooling conditions, which are of the greatest importance to the method of manufacture of the present invention are described. The steel must be cooled to the higher of 600° C at a mean cooling rate of not less than 5° C/second from the A_{e3} transformation point, or alternatively to a temperature T_1 which satisfies formula (1), which is the temperature at which the ferrite transformation is initiated under the normal hot rolling process, which is to say the temperature at which is formed a grain boundary ferrite structure from the austenite grain boundaries and the transformation zone, and when an intragranular ferrite microstructure is formed in the final transformation as envisaged by the present invention, such ferrite that is formed from the grain boundaries may possess a coarser structure and may be irregular.

Moreover, the temperature at which intragranular ferrite is normally formed is close to T_1 , and that the rapid cooling employed heretofore has the effect of amplifying the driving force of the ferrite transformation. However, when such temperature exceeds 600° C, a bainite structure is formed, and hence direct cooling from the austenite region at up to 600° C is desirable.

Thus, after the steel has been cooled to the temperature limits envisaged by the present invention, the steel is held for several minutes in the vicinity of such temperature, or is further cooled to a lower temperature, whereby the intragranular ferrite transformation may

readily be completed. While no particular restriction is applied, when working of the steel is performed in the austenite region, intragranular ferrite is more readily formed than when the steel is in the unworked state, and hence a faster rate of cooling than that described here is desirable.

5 Next, the steel is again reheated from this state to form austenite, but at this point, the transformation to the initial intragranular ferrite structure must be complete. This is because, if any austenite structure remains when the transformation to austenite commences, the formation of new austenite nuclei is rendered more difficult.

10 For similar reasons, when performing the transformation to austenite, reheating should preferably not be performed until the intragranular ferrite structures have been completely extinguished. While no particular restrictions apply, in order to achieve the finest possible microstructure when the austenite transformation has been completed, the rate of increase in the temperature should preferably be high, and the final temperature and the retention time at the conclusion of reheating should be as low and as short as possible.

15 The fine-grained austenite structure which is obtained by this process is fundamentally no different from the structure which is obtained as a result of repeated recrystallization in the conventional rolling process. Thus the surface area of the austenite grain boundaries at this time is virtually no different from that observed in steels manufactured by the conventional process. Thus, by means of the present invention, if after the completion of such transformation to austenite rolling is performed as a repeated form of secondary working,
20 recrystallization produces a finer grain structure and dislocation and deformation zones are introduced, the ferrite transformation sites in the succeeding secondary cooling may be increased.

Moreover, the conditions for such secondary cooling are not materially different from those of the conventional hot rolling process. Thus, by the selection of cooling conditions and coiling conditions which are appropriate to the circumstances, and provided that the conditions for the composition of the steel as envisaged by the present invention are observed, a fine-grained polygonal ferrite structure may be obtained, and ferrite structures and mixed structures consisting of bainite, martensite and pearlite and so forth, may be obtained in the same manner as with the conventional process, and hence steels which meet various requirements for mechanical properties may be manufactured.

Practical embodiments

Practical Embodiments 1

Steel consisting of 0.10 wt% C, 0.50 wt% Si, 1.42 wt% Mn, 0.008 wt% P, 0.011 wt% S, 0.034 wt% Al, and with the balance iron and inevitable impurities is formed in a converter, and is continuously cast to form thin cast ingots of from 3 mm to 50 mm. The A_{e3} transformation point which is calculated from such components is 839°C , and the temperature T_1 that is calculated from formula (1) is 639°C .

Next, such thin cast ingots rolled, cooled and reheated according to the conditions set out in Table 1 and through the thermal history illustrated in Figure 1, and the thin cast ingots are formed into sheet having a thickness of 3 mm. In the thermal history of the steel, the primary working temperature of the steel is 1000°C , the rate of temperature increase in order to achieve the austenite transformation is 5°C/second , no secondary working is performed, the secondary cooling conditions are cooling from 900°C to 600°C at a rate of 7°C/second , and the sheet is coiled at a temperature of 600°C . Moreover the steel is held at the initial cooling stop temperature T_c , and at the reheating end temperature T_h , for 2 minutes for each.

Table 1

No.	Rate of reduction (%)	CR (° C/s)	Tc (° C)	Th (° C)	Remarks
1	0	10	600	1000	Present invention
2	20	<u>2</u>	550	950	Comparative example
3	70	5	350	<u>700</u>	Comparative example
4	15	5	400	900	Present invention
5	30	10	<u>700</u>	1000	Comparative example
6	60	15	550	1025	Present invention
7	90	10	550	1000	Comparative example
8	0	7	600	<u>800</u>	Comparative example

Values underlined are outside the range envisaged by the present invention.

$Ae_3 = 839^\circ \text{C}$ (Ae_3 calculated by means of THermo-CALC Ver. F)

$T_1 = 639^\circ \text{C}$

The steel sheets are pickled and are then subjected to the following mechanical tests. 1) Samples are prepared according to JIS Z 2201.5 and each sample is subjected to tensile testing according to the method of JIS 2241, and the yield strength, tensile strength, and total elongation of the samples are measured. 2) Samples are prepared according to JIS Z 2202.4 (2.5 mm) and each sample is subjected to Charpy impact testing according to the method of JIS 2242, and the surface transition temperatures of the samples are measured. 3) Samples are prepared according to JIS Z 2275.1 and each sample is subjected to fatigue testing according to that method, and the fatigue strengths of the samples are measured. 4) Samples are prepared by punching holes having initial diameters of 10 mm in order to evaluate the stretch flanging characteristics of the samples, and the hole expansion ratio, which is the

value calculated by dividing the average hole diameter at the point when a crack penetrates the width of the sheet by the initial hole diameter, is calculated.

Table 2 lists the results of examinations by optical microscope of the microstructures of the samples, and the results of the mechanical tests.

Table 2

No.	Microstructure	Yield point	Tensile strength	Elongation	Strength/ductility balance	Transition temperature	Fatigue strength	Fatigue limit	Hole expansion ratio	Remarks
		YP (kgf/mm ²)	TS (kgf/mm ²)	El (%)	TS x El	VTIS (°C)	σ_y (kgf/mm ²)	σ_y/TS	d/d_0	
1	Fine grained, regular PF	41.2	63.0	29.6	1865	-56	33.4	0.53	2.3	Present invention
2	Irregular PF + VF	33.4	58.3	17.9	1044	-13	24.5	0.42	1.8	Comparative example
3	Irregular VF:	54.1	69.1	19.6	1354	+7	24.9	0.36	1.4	Comparative example
4	Fine grained, regular PF	40.6	62.5	30.4	1900	-61	38.1	0.61	2.4	Present invention
5	Fine grained PF	57.2	70.6	20.5	1547	-52	28.2	0.40	1.7	Comparative example
6	Fine grained, regular PF	39.1	60.2	31.7	1908	-54	34.3	0.57	2.0	Present invention
7	Coarse grained PF	35.0	42.0	32.8	1705	-19	52.4	0.38	1.5	Comparative example
8	Irregular, coarse-grained PF	32.8	54.7	28.3	1548	-31	15.9	0.29	1.2	Comparative example

WF: Acicular Widmanstätten ferrite

PF: Polygonal ferrite

Specimen Numbers 1, 4 and 6 which were prepared by the method envisaged by the present invention possess structures of fine-grained regular polygonal ferrite and hence possess high strength-ductility balances of not less than 1800, toughness of not more than -50°C for vTrs, fatigue characteristics of fatigue limits of not less than 0.5, and hole extension characteristics of hole extension ratios of not less than 2.0, which are the maximum levels for these characteristics for steel sheet which may be obtained by manufacture by the conventional hot rolling process.

On the other hand, the rate of cooling at the primary cooling for Specimen No.2 was low and hence a regular intragranular ferrite structure could not be obtained, and such deficiency could not be overcome by subsequent reheating and secondary cooling, such that the strength-ductility balance and hole extension characteristics of the steel were degraded.

Moreover, the rate of reduction for Specimen No.7 exceeded 80% and hence the microstructure obtained at primary cooling was principally of grain boundary ferrite rather than of intragranular ferrite, and it is believed that, as a result of the growth in the austenite ratio from the incomplete initial structure, the austenite also became coarse-grained, such that a fine-grained final microstructure could not be obtained. For this reason, the toughness and fatigue characteristics in particular of the steel were degraded.

Moreover, the conditions for the upper and lower temperature limits at primary cooling and reheating were not observed in the case of Specimens Nos. 3, 5, and 8, and hence the ferrite transformations in each case were not completed, a single austenite phase was not achieved, and the microstructures of these specimens were not regular. In the case of Specimen No.5 in particular, the effects of reheating were in no way reflected in the structure of the steel, and the characteristic values for this steel match those for an intragranular ferrite structure.

Thus such steel possessed excellent strength and toughness, but poor elongation and hole extension characteristics.

Practical Embodiment 2

5 Steels having the chemical compositions listed in Table 3 were prepared in a converter and continuously cast to form 6 mm thin steel strip, whereupon such steels were subjected to primary working in the austenite region at 1000° C and were subjected to 50% reduction to form 3 mm cast ingots. Next, the specimens were subjected to primary cooling during which they were cooled from 950° C to 500° C at a rate of 5° C/second; the steels were held for 3 minutes at 500° C and were then reheated to 1000° C at a rate of 10° C/second, and then
10 were immediately subjected to secondary cooling during which they were cooled to 600° C at a rate of 20° C/second, and the specimens were then coiled at that temperature. Subsequently the specimens were subjected to the same observations of structure and mechanical tests as were employed for Practical Embodiments 1.

The results are set out in Table 4.

Table 3

No	Chemical composition (wt%)								Ae ₃ (°C)	T ₁ (°C)	Remarks
	C	Si	Mn	P	S	Al	N	Other			
1	0.080	0.20	1.52	0.009	0.008	0.032	0.0021		831	671	Present invention
2	0.023	0.03	<u>0.20</u>	0.016	0.014	0.045	0.0032		894	848	Comparative example
3	0.126	0.45	1.00	0.007	0.005	0.025	0.0020	Ti: 0.015; B: 0.0010	844	592	Present invention
4	<u>0.003</u>	0.05	<u>0.10</u>	0.010	0.010	0.042	0.0038	Nb: 0.005	909	903	Comparative example
5	0.015	0.21	0.98	0.008	0.012	0.016	0.0018	Mo: 0.01	864	834	Present invention
6	0.153	1.27	1.37	0.012	0.008	0.031	0.0034	Ni: 0.4; Cr: 0.5	835	529	Present invention
7	0.056	0.09	1.42	0.008	0.015	0.022	0.0025	V: 0.05	837	725	Present invention
8	0.091	0.03	1.50	0.012	<u>0.002</u>	0.035	0.0016		823	641	Comparative example

Values underlined are outside the range envisaged by the present invention.

Ae₃ calculated by means of THermo-CALC Ver. F

Table 4

No.	Microstructure	Yield point	Tensile strength	Elongation	Strength-ductility balance	Transition temperature	Fatigue strength	Fatigue limit	Hole expansion ratio	Remarks
		YP (kgf/mm ²)	TS (kgf/mm ²)	El (%)	TS x El	VTIS (°C)	σ_f (kgf/mm ²)	σ_f/TS	d/d ₀	
1	Fine grained, regular PF	37.8	63.2	28.7	1813	-52	34.1	0.54	2.2	Present invention
2	Irregular PF + bainite	48.0	56.0	19.6	1098	-34	24.1	0.43	1.8	Comparative example
3	Fine grained, regular PF	46.8	72.3	25.0	1807	-65	43.4	0.60	2.2	Present invention
4	Coarse grained PF	21.6	36.1	44.3	1599	-31	12.6	0.35	2.0	Comparative example
5	Fine grained, regular PF	35.4	50.6	33.8	1710	-72	28.8	0.57	2.3	Present invention
6	Fine grained, regular PF	56.3	78.2	28.0	2190	-60	41.4	0.53	2.1	Present invention
7	Fine grained, regular PF	42.2	65.0	26.2	1625	-50	35.1	0.54	2.4	Present invention
8	Irregular PF + martensite	73.7	83.7	15.4	1353	-7	21.8	0.26	1.3	Comparative example

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W.F: Acicular Widmanstätten ferrite

P.F: Polygonal ferrite

Specimen Nos. 1, 3, 5, 6, and 7 possessed compositions according to the conditions of the present invention, and possessed fine-grained regular polygonal ferrite microstructures, and hence exhibited high strength-ductility balances of not less than 1600, toughness of not more than -50°C for vTrs, fatigue characteristics such as fatigue limits of not less than 0.5, and hole extension characteristics such as hole extension ratios of not less than 2.0, which are the maximum levels for these characteristics for steel sheet which may be obtained by manufacture by the conventional hot rolling process.

On the other hand, the compositions of Specimens Nos. 2 and 8 contained levels of Mn and S which were outside the conditions of the present invention, and hence the precipitation of MnS and so forth which form the nuclei for the intragranular ferrite transformation prior to the primary cooling was inadequate, and hence the initial microstructure prior to reverse transformation was irregular, and such irregularity was sustained through into the final microstructure. For these reasons, the elongation and hole extension characteristics of these Specimens were degraded.

Specimen No.4 contained less C than the level envisaged by the present invention, and hence the initial microstructure obtained was not fully intragranular ferrite, and as a result, the final microstructure was coarse-grained and the fatigue limit of the specimen was degraded.

Effects of the invention

By means of the present invention, the steel ingot may be employed in an as-cast condition after continuous casting or may be subjected to light rolling, whereby the conventional hot rolling process may be omitted or reduced, and a regular fine-grained polygonal ferrite microstructure similar to that obtained with the conventional hot rolling process may be obtained, such that thin steel sheet which possesses excellent mechanical properties such as strength-ductility balance, toughness, fatigue characteristics, hole extension characteristics

and so forth, and which are comparable with those mechanical characteristics obtainable by means of the conventional hot rolling process, may be obtained.

Moreover, steel types equivalent to the conventional types of steel may be manufactured by new processes at the steel production and rolling stages, and hence considerable increases in productivity and reductions in costs may be achieved.

Moreover, steel sheet manufactured according to the present invention never exhibits the banded irregular microstructures which are commonly formed in conventional hot rolled materials, and the present invention forms completely isotropic structures and hence such steel sheet forms an excellent material for applications such as bending and press forming in which anisotropy would be a problem.

Moreover, the present invention is intended principally to provide steel sheet having gauges comparable with those of hot rolled steel sheet, but the scope of the application of the invention may be extended to processes for the manufacture of plate of 30 mm gauge from thin cast ingots of 50 mm in thickness, through the application of the same fundamental metallurgical phenomena as are envisaged by the present invention.

4. Simplified Description of the Diagram

Figure 1 is a figure which represents the thermal history after casting as envisaged by the present invention.

Agent: Tateo Sanoki, Patent Attorney

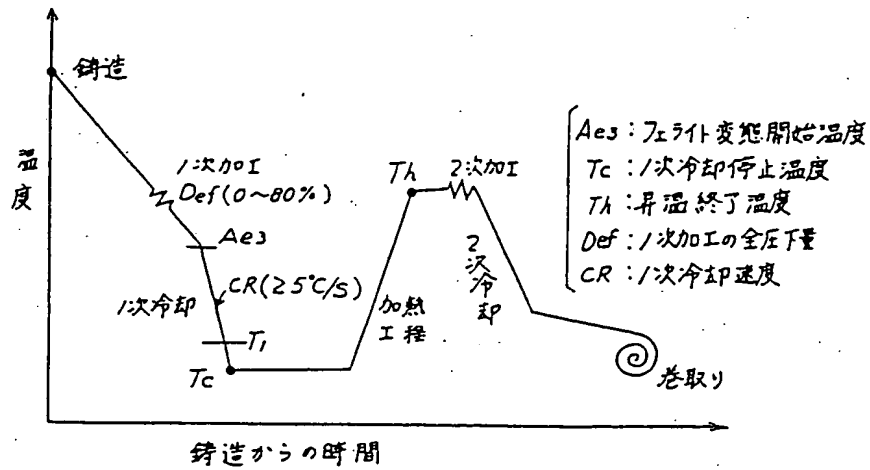


Figure 1

A: Temperature; B: Time after casting; C: Casting; D: Primary working; E: Primary cooling;
 F: Reheating process; G: Secondary working; H: Secondary cooling; I: Coiling; A_{e3} : Ferrite
 transformation initiation temperature; T_c : Primary cooling stop temperature; T_h : Heating
 stop temperature; Def: Total reduction in primary working; CR: Primary cooling rate

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⑮ 発明の名称 薄帯鋼板の製造方法

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明 細 書

1. 発明の名称

薄帯鋼板の製造方法

2. 特許請求の範囲

1. 重量比で、

C : 0.01~0.2%、

Si : 3.0%以下、

Mn : 0.5~3.0%、

P : 0.10%以下、

S : 0.003~0.05%、

Al : 0.01~0.1%

残部鉄及び不可避免的不純物よりなる溶鋼を連続
鋳造にて鋳造後、 A_{e3} 変態点以上のオーステナ
イト域で無加工もしくは全圧下量で80%以下の
一次加工を施した後、 A_{e3} 変態点から平均冷速
5℃/s以上で600℃もしくは(1)式を満たす温
度 T_1 の内どちらか高い方の温度以下まで一次冷
却し、粒内フェライト組織を形成せしめ、変態が
完全に完了した後再び A_{e3} 変態点以上まで加熱

しオーステナイト単相組織とし、再度二次冷却す
ることで得られる、異方性が少なくなかつ微細
なポリゴナルフェライト組織よりなることを特徴
とする薄帯鋼板の製造方法。

$$T_1 = A_{e3} - 2000 \times C \text{ (wt\%)} \quad (1)$$

2. 通常の成分としてNb, Ti, V, Ni, Cr, Mo, Bの内1種または2種以上を1.0%以下含有する鋼を用いる特許請求の範囲第1項記載の薄帯鋼板の製造方法。

3. 発明の詳細な説明

(産業上の利用分野)

本発明は鋼板の製造方法に関し、詳しくは、熱間圧延工程を省略もしくは簡略化して、連続鋳造鋳片から直接に、強度・延性バランス等の従来の熱延鋼板で必要とされる機械的性質の優れた鋼板を製造する方法に関するものである。

(従来の技術)

近年、連続鋳造機の著しい開発により、鋳片の薄手化が図られ、従来の熱延工程の省略、簡略化が進みつつある。またこのような薄手化は熱延工

程に於けるコスト低減の観点から注目されている。

しかしながら、最終板厚に近い鋳片からの製造工程では、従来の熱延工程のような大きな圧下量が取れないため、一般にはフェライト組織が粗大化し、十分な機械的性質を満足し得ないことが問題点として指摘されている。ちなみに従来工程に於ける熱延圧下率は最低80%とされている（橋本嘉男：鉄と鋼72（1986），p.2283）。

上記の組織の粗大化の原因は、オーステナイトからフェライトに変態する際、オーステナイト組織が粗大なままで、その粒界面積が少なく、また変形帯などの欠陥が少ないため、フェライトの核生成サイトが少ないことによる。

このような問題点に対し、近年、酸化物などをフェライトの変態核としてオーステナイトの粒内からも積極的に変態を行わせる粒内フェライトの活用が厚板鋼板を中心に検討されている（例えば特開昭61・213822号公報）。

但し、この方法はその核となる酸化物や析出物を均一分散させるなどの組織制御の点から成分管

(100)集合組織をいかにランダム化するかに焦点が絞られており、熱延終了時相当における、鋼板の機械的性質については何も言及されていない。（発明が解決しようとする課題）

熱間圧延工程を省略もしくは簡略化して、逆統銑造鋳片から直接に、強度・延性バランス等の従来の熱延鋼板で必要とされる機械的性質の優れた鋼板を製造する方法を確立することが、本発明の目的である。

（課題を解決するための手段）

強度・延性バランスや疲労特性などの優れた機械的性質を有する熱延鋼板に共通して見られる金属組織学的特徴は、組織が微細であり、均一であり、そして一つ一つの結晶粒がポリゴナルであることである。すなわち薄鋳片から出発するような簡略プロセスにおいても、最終的にオーステナイトから変態した組織が微細なポリゴナルフェライト主体となれば必要とする機械的性質を満たすこととなる。

本発明者らは、上記の実情に鑑み鋭意検討した

理や溶解条件が難しく、またオーステナイト粒界から生成するフェライトサイドプレートやベイナイトなどが変態時に競合するため均一な組織になりにくい。また粒内フェライト組織そのものが極めて針状のウィッドマンシュテッテンフェライトになり易く、靱性に対しては極めて有効なものの、延性や疲労特性などの機械的性質に対してはあまりその効果はない。またこの技術の最大の問題点はオーステナイト時の圧下によって上記の組織の不均一化が助長されることにあり、局部的な現象が重要な延性などには極めて不利である。

一方、薄鋳片に於ける組織の粗大化に対する、別のアプローチとしては特開昭61・99630号や特開昭63・62822号公報などに開示されているインラインでの再熱処理法が挙げられる。但しこれらはプレス成形性に優れた深絞り用冷延鋼板を前提としており、いわば冷延素材を得るための熱延代替プロセスとして検討されたものである。

そのため、再熱処理によるフェライトからオーステナイトへの逆変態の効果も鋳造組織に特有の

結果、従来技術として検討されている粒内フェライト変態のみによっては完全なポリゴナルフェライト組織のみを生成させることが不可能であるとの結論に達した。

また逆変態を付加するプロセスにおいても、その繰り返し熱処理による細粒化効果は前組織の影響を強く受け継ぎ、特に前組織が通常の凝固組織のような粗大なフェライト組織では、わずか一回の逆変態によりオーステナイト化させ再度二次冷却によりフェライト変態させても、その粒径の変化はそれほどないことを確認した。

この後者の現象については特開昭63・115654号公報においても確認されており、発明者らは繰り返し熱処理を2回以上することが必要としている。

しかしながら、本発明者らはこのような種々の問題点を回避して、わずか一回の逆変態を付加するのみで微細なポリゴナルフェライト組織を得るプロセスを種々の実験により見いだしたのである。

その基本となる原理は、第1図に示すような熱履歴でフェライトからオーステナイトへの逆変態

を行う際、その初期フェライト組織を粒内フェライト組織にすることにある。

すなわち本発明は、成分及び初期フェライト変態の際の一次冷却条件を制御することで、凝固からの粗大なオーステナイト組織を粒内フェライトにより一旦微細組織としておき、その後、逆変態を行わせることにより、微細なオーステナイト組織を形成せしめるところにその特徴があり、この結果、本発明によれば最終的な二次冷却により得られるフェライト組織が微細となり、なおかつ組織が不均一化、ポリゴナル化する。

このメカニズムの詳細はまだ不明であるが、粒内フェライト組織がオーステナイト変態の際の極めて多数のオーステナイトの核生成サイトを有していること、また粒内フェライトの方位が各粒で異なるため、逆変態の際、各々のフェライト粒界から核生成したオーステナイト粒もまた方位が異なり、オーステナイト化終了時においても微細粒を保つためと考えられる。

また本発明によれば、逆変態を付加することに

より、従来の粒内フェライト組織で問題となっていた組織の不均一化を解消することができるため、この不均一化を軽減するため製造上限定されていた、従来の成分条件、圧下条件などの緩和を図ることができる。

つまり、本発明はC: 0.01~0.2%、Si: 3.0%以下、Mn: 0.5~3.0%、P: 0.10%以下、S: 0.003~0.05%、Al: 0.01~0.1%を含み、残部鉄及び不可避的不純物よりなる溶鋼を連続鋳造にて鋳造後、 A_{e_3} 変態点以上のオーステナイト域で無加工もしくは全圧下量で80%以下の一次加工を施した後、 A_{e_3} 変態点から平均冷速 $5^{\circ}\text{C}/\text{s}$ 以上で 600°C もしくは(1)式を満たす温度 T_1 の内どちらか高い方の温度以下まで一次冷却し、粒内フェライト組織を形成せしめ、変態が完全に完了した後再び A_{e_3} 変態点以上まで加熱しオーステナイト単相組織とし、再度二次冷却することで得られる、異方性が少なくなおかつ微細なポリゴナルフェライト組織よりなる、強度・延性バランス、靱性、疲労特性、穴広げ性が優れた薄帯

鋼板の製造方法である。

$$T_1 = A_{e_3} - 2000 \times C \quad (\text{v}\%) \quad (1)$$

又通常の成分としてNb, Ti, V, Ni, Cr, Mo, Bの内1種または2種以上を1.0%以下含有する鋼を用いることができる。

まず、本発明における化学成分の限定理由について述べる。

Cはオーステナイトからフェライトへの変態において、その組織形態を決定する最も重要な元素であり、本発明のように初期の変態組織としてウィッドマンシュテッテンフェライトを安定に得るためには、その下限を0.01%とする。またその上限としては溶接性を劣化させることのないよう0.2%とする。

Siは鋼の強度・延性バランスを劣化させずに高強度化するために添加すると共に、変態点を上げてフェライトの析出を促進させる働きを持つ。しかし過度の添加は延性及び表面性状の劣化をきたすため、その上限を3.0%とする。

MnはCと同様、鋼の高強度化を目的に添加さ

れる元素であると共に、粒内フェライト組織形成のために必要不可欠な元素である。すなわちこれは後述するSと鋼中でMnSを形成し、Al, SiやMnなどの鋼中の酸化物構成元素より形成される酸化物上に優先的に析出することにより、有効なフェライトの析出核となり、粒内フェライト変態を促進させる。このため添加量の下限はこの効果が明瞭に現れる0.5%とし、また上限については製鋼工程における組成制御のコストが低く抑えられ、また加工性を劣化させることのないように3.0%とする。

Pも鋼を高強度化するために添加する元素であるが、過度の添加は延性及び溶接性を劣化させるため、その上限を0.10%とする。

SはMnと同様、本発明において重要な役割を果たすMnSの構成元素である。しかし過度の添加は熱間割れの原因となるため、その上限を0.05%とし、また下限については脱酸コストの上昇などの問題から0.003%とする。

Alは鋼の脱酸のために必要であり、本発明に

においてはその下限を0.01%とする。一方、過剰の添加はコストの上昇を招くと共に、鋼中に介在物として残留し割れなどの原因となることからその上限は0.1%とする。

Nb, Ti, V, Ni, Cr, Mo, Bは鋼の強度を高めるなどの目的で添加される。特にTi, Vなどは粒内フェライト変態促進元素としても高い効果がある。このため本発明においては、これらの元素の内1種または2種以上を1.0%以下の範囲で添加してもよい。

次に製造方法について述べる。

本発明においては、以上述べたような成分の鋼を連続鋳造後、鋳造ままもしくは A_{e_3} 変態点以上の温度において全圧下量で80%以下の一次加工を施した後、最初のフェライト変態を行う。ここで加工量に上限を設けたのは、これ以上の圧下を行うことは従来の熱延工程と冶金組織学的にも、また設備コスト及び製造コストの上でも何ら差異がなくなるためである。一方、鋳片の表面性状を整えるなどの目的で行われる軽圧下は、すべてこ

の圧下量に含まれる。

次に本発明で最も重要な一次冷却条件は、 A_{e_3} 変態点から平均冷速 $5^\circ\text{C}/\text{s}$ 以上で600℃もしくは(1)式を満たす温度 T_1 の内どちらか高い方の温度以下まで冷却しなければならない。なぜならば上記の A_{e_3} から600℃もしくは T_1 までの温度範囲は通常の熱延工程におけるフェライト変態開始温度、すなわちオーステナイト粒界や変形帯などからのいわゆる粒界フェライトと呼ばれる組織の成形温度にあたり、本発明のように最初の変態において粒内フェライト組織とする場合、これらの粒界からのフェライト生成は組織の粗大化、および不均一化を招くからである。

また通常、粒内フェライトの生成温度は T_1 近傍であり、ここまでの急冷はフェライト変態の駆動力を増大させる効果を持つ。ただしこの温度が600℃を超える場合はベイナイト組織の生成が起るため、600℃以下までのオーステナイト域からの直接の急冷は好ましくない。

さて、本発明法によればこれらの限定温度まで

冷却後、その温度近傍で数分間保持するか、さらに低温まで冷却することによって容易に粒内フェライト変態を完了させることができる。なお特に限定はしないがオーステナイト域で加工を行った場合、無加工状態と比較して粒界フェライトが生成し易くなるため、ここでの冷却速度はより高めが望ましい。

つづいて本発明ではこの状態から再度加熱しオーステナイト化するが、その際初期の粒内フェライト組織への変態は完全に終了していなければならない。これはオーステナイト化の初期に残留のオーステナイト組織があると、新たなオーステナイトの核生成が行われにくくなるためである。

また同等の理由により、オーステナイト化する際にも完全に粒内フェライト組織が消滅するまで加熱することが望ましい。ただし限定はしないが、オーステナイト化終了時にできるだけ微細組織とするためには、その昇温速度は大きく、また加熱終了温度やその温度での保定等もできるだけ低くまた短くすることが望ましい。

このような工程で得られた微細なオーステナイト組織は、従来の圧延による再結晶を繰り返した結果得られる組織と基本的な差異はない。すなわちこの時点でオーステナイトの粒界面積は従来工程材とほとんど変わりはない。よって本発明においてもこのオーステナイト化終了後、再度二次加工として圧延を施し、再結晶によりさらなる微細化をしたり転位や変形帯を導入して、続く二次冷却時のフェライト変態のサイトを増大させることも可能である。

またこの二次冷却条件も特に従来熱延工程と異なるものではない。すなわちここでの冷却条件および巻取り条件を適宜選ぶことにより、本発明の成分条件内で、微細なポリゴナルフェライト組織のみとすることはもちろん、フェライトとベイナイトやマルテンサイトやパーライトなどとの混合組織を従来工程同様作り分けることが可能となり、各種の機械的性質を満足する鋼板の製造が可能となる。

(実施例)

実施例 1

重量%で、C:0.10、Si:0.50、Mn:1.42、P:0.008、S:0.011、Al:0.034、残部Fe及び不可避免的不純物からなる鋼を転炉出鋼し、連続铸造にて3~50mmの薄鋼片とした。なおこの成分から推定される A_{e3} 温度は839℃、また(1)式から計算される T_1 温度は639℃である。

続いて第1図に示すような熱履歴で、第1表に示すような圧下・冷却及び加熱を行い、鋼片の板厚を3mmとした。熱履歴の内、一次加工温度は1000℃、オーステナイト化のための昇温速度は5℃/s、二次加工は行わず、二次冷却条件は900~600℃を7℃/sとし、600℃で巻取った。また最初の冷却停止温度 T_c 及び加熱終了温度 T_h でそれぞれ2minの保温を行った。

鋼板は酸洗後、以下の機械試験を行った。1) JIS Z 2201.5号試験片に加工し、同2241記載の試験方法に従って引張試験に供し、降伏強度・引張強度・全伸びを測定した。2) JIS Z 2202.4号試

験片(2.5mm)に加工し、同2242記載の試験方法に従ってシャルピー衝撃試験に供し、破面遷移温度を測定した。3) JIS Z 2275.1号試験片に加工し、記載の試験方法に従って疲労試験に供し、両張り引張時の疲労強度を測定した。4) 伸びフランジ性の評価として初期穴径10mmでの打ち抜き穴広げ試験に供し、クラックが板厚を貫通した時点での平均穴径を初期穴径で除した値、いわゆる穴広げ比を測定した。

第2表に光学顕微鏡組織観察結果およびこれらの機械試験の結果を示す。

第 1 表

No	圧 下 率 (%)	CR (℃/s)	T_c (℃)	T_h (℃)	備 考
1	0	10	600	1000	本発明
2	20	2	550	950	比 較
3	70	5	350	700	比 較
4	15	5	400	900	本発明
5	30	10	700	1000	比 較
6	60	15	550	1025	本発明
7	90	10	550	1000	比 較
8	0	7	600	800	比 較

下線の値は本発明の範囲外を示す。

$A_{e3} = 839^\circ\text{C}$ (A_{e3} はTHERMO-CALC Ver.Fにより計算)
 $T_1 = 639^\circ\text{C}$

第 2 表

No	ミ ク ロ 組 織	降 伏 点	引 張 強 度	全 伸 び	強 度 ・ 延 性 バ ラ ン ス	遷 移 温 度	疲 勞 強 度	疲 勞 限	穴 広 げ 比	備 考
		Y P (kgf/mm ²)	T S (kgf/mm ²)	E l (%)	T S × E l	v T R s (℃)	σ_v (kgf/mm ²)	$\sigma_v / T S$	d / d ₀	
1	微細、均一 PF	41.2	63.0	29.6	1865	-56	33.4	0.53	2.3	本発明
2	不均一 PF+WF	33.4	58.3	17.9	1044	-13	24.5	0.42	1.8	比 較
3	不均一 WF	54.1	69.1	19.6	1354	+7	24.9	0.36	1.4	比 較
4	微細、均一 PF	40.6	62.5	30.4	1900	-61	38.1	0.61	2.4	本発明
5	微細 WF	57.2	70.6	20.5	1547	-52	28.2	0.40	1.7	比 較
6	微細、均一 PF	39.1	60.2	31.7	1908	-54	34.3	0.57	2.0	本発明
7	粗大 PF	35.0	52.0	32.8	1705	-19	32.4	0.38	1.5	比 較
8	不均一、粗大 PF	32.8	54.7	28.3	1548	-31	15.9	0.29	1.2	比 較

WF: 針状ワイドマンシュテンフェライト
 PF: ポリゴナルフェライト

本発明条件の№1, 4, 6では組織が微細で均一なポリゴナルフェライトよりなっているため、強度・延性バランスは1800以上と高く、また靱性も vT_{rs} で -50°C 以下、疲労特性も疲労限で0.5以上、穴広げ特性も穴広げ比で2.0以上といった、この成分系を用いて従来熱延工程で製造される鋼板の最高レベルの特性が得られている。

一方、№2は一次冷却時の冷速が遅かったため、均一な粒内フェライト組織が得られず、これがその後の再加熱・二次冷却によっても解消されなかったために、強度・延性バランス、及び穴広げ比が劣っている。

また№7は圧下量が80%を超えたため、一次冷却時の組織が粒内フェライト主体と言うよりむしろ粒界フェライト主体となってしまっており、中途半端な初期組織からオーステナイト比が進行した結果、オーステナイトも粗粒となり、最終組織も微細にならなかったものと思われる。このため特に靱性、疲労特性の点で劣っている。

さらに№3, 5, 8は一次冷却及び加熱時にそ

の下限、上限温度が本発明条件から外れていたため、それぞれフェライト変態が終了していなかったり、オーステナイト単相組織まで至らず、組織が不均一化している。特に№5では再熱の効果が全く組織に反映されておらず、この材料の特性値は粒内フェライト組織そのものの持つ特性と一致している。すなわち強度、靱性などは優れているものの、伸び、穴広げ特性は劣っている。

実施例 2

第3表に示した化学成分の鋼を転炉出鋼し、連続鋳造にて6mmの薄鋳帯に鋳造後、オーステナイト域の1000℃で一次加工として50%の圧下を加えて3mmの鋳片とした。続いて一次冷却として950～500℃までを15℃/sで冷却し、500℃で3min保た後、10℃/sで1000℃まで加熱し、ただちに20℃/sで600℃までの二次冷却を施し、その温度で巻取った。その後、実施例1と同じ方法にて組織観察及び機械試験を行った。

第4表にその結果を示す。

第 3 表

No	化 学 組 成 (wt%)							Acs (℃)	T1 (℃)	備 考
	C	Si	Mn	P	S	Al	N			
1	0.080	0.20	1.52	0.009	0.008	0.082	0.0021	891	671	本発明
2	0.023	0.03	0.20	0.016	0.014	0.045	0.0032	894	848	比 較
3	0.126	0.45	1.00	0.007	0.005	0.025	0.0020	844	592	本発明
4	0.003	0.05	0.10	0.010	0.010	0.042	0.0038	909	908	比 較
5	0.015	0.21	0.98	0.008	0.012	0.018	0.0018	864	834	本発明
6	0.153	1.27	1.37	0.012	0.008	0.081	0.0034	885	529	本発明
7	0.056	0.09	1.42	0.008	0.015	0.022	0.0025	837	725	本発明
8	0.091	0.03	1.50	0.012	0.002	0.035	0.0016	828	641	比 較

下線の値は本発明の範囲外を示す。

(Acs は THERMO-CALC Ver.F により計算)

第 4 表

No	ミクロ組織	降伏点	引張強度	全伸び	強度・延性 バランス	遷移温度	疲労強度	疲労限	穴広げ比	備 考
		YP (kgf/mm ²)	TS (kgf/mm ²)	El (%)	TS×El	vT_{rs} (℃)	σ_v (kgf/mm ²)	σ_v/TS	d/d ₀	
1	微細、均一PF	37.8	63.2	28.7	1813	-52	34.1	0.54	2.2	本発明
2	不均一PF+ベイナイト	48.0	58.0	19.6	1098	-34	24.1	0.43	1.8	比 較
3	微細、均一PF	46.8	72.3	25.0	1807	-65	43.4	0.60	2.2	本発明
4	粗大PF	21.6	36.1	44.3	1599	-31	12.6	0.35	2.0	比 較
5	微細、均一PF	35.4	50.6	33.8	1710	-72	28.8	0.57	2.3	本発明
6	微細、均一PF	58.3	78.2	28.0	2190	-60	41.4	0.53	2.1	本発明
7	微細、均一PF	42.2	65.0	26.2	1625	-50	35.1	0.54	2.4	本発明
8	不均一PF+マルテンサイト	73.7	83.7	15.4	1353	-7	21.8	0.26	1.3	比 較

(WF: 針状ワイドマンシュテンフェライト
PF: ポリゴナルフェライト)

本発明条件の成分である№1, 3, 5, 6, 7は組織が微細で均一なポリゴナルフェライトよりなっているため、強度・延性バランスは1600以上と高く、また靱性も vT_{rs} で -50°C 以下、疲労特性も疲労限で0.5以上、穴広げ特性も穴広げ比で2.0以上といった、それぞれの成分系を用いて従来熱延工程で製造される鋼板の最高レベルの特性が得られている。

一方、№2, 8はそれぞれMn及びS量が本発明条件から外れているために、一次冷却前に粒内フェライト変態の核となるMnSなどの析出が十分行われず、その結果逆変態の前の初期組織が不均一になって、最終組織にまでその不均一性が持ちこたされている。このため特に伸び、穴広げ特性が劣っている。

また№4はC量が低く本発明条件から外れているためやはり初期変態組織として粒内フェライト組織とならず、その結果最終組織も粗大となって疲労限などが劣っている。

対して適用されるものであるが、本発明の基本的冶金現象を用いれば厚み50mm程度の薄鋼片から30mm程度の厚板材を製造するような工程に対して、その適用範囲は広げることが可能である。

4. 図面の簡単な説明

第1図は本発明の製造後の熱履歴を表した図表である。

代理人 弁理士 茶野木 立 夫

(発明の効果)

本発明によれば、鋼片を連続製造後そのまま、もしくは軽圧下を施すのみの熱延簡略もしくは省略プロセスにおいても、従来熱延工程と同様の均一かつ微細なポリゴナルフェライト組織とすることが可能であり、その結果、従来の熱延鋼板が有する強度・延性バランス、靱性、疲労特性、穴広げ性などの種々の機械的性質の優れた薄帯鋼板の製造が可能となる。

またこのような製鋼・圧延段階での新プロセスにおいても従来同等の鋼種の作り分けができることから、飛躍的な生産性の向上と設備コストの低減が可能となる。

さらに本発明により製造した鋼板は、しばしば通常の熱延材において圧延により生じる、バンド状の不均一組織が全く見られないため、完全に等方的な組織として、異方性が問題となるような用途、例えば曲げ加工やプレス成形性に対しても極めて優れた材料と言える。

また本発明は主として熱延鋼板相当の板厚材に

第1図

